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THE FRACTURE ENERGY OF COMPOSITE MATERIALS

Final Report

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Burlington, Vermont
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THE FRACTURE ENERGY OF COMPOSITE MATERIALS

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FINAL REPORT

(September 1, 1966 to August 31, 1967)

By:

John O. Outwater and William O. Carnes

Submitted to:

U. S. Army Munitions Command
Picatinny Arsenal
Dover, New Jersey

September 30, 1967

Approved by: _____

J. O. Outwater
Principal Investigator

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FOREWORD

This report has been prepared by the University of Vermont under Contract No. DAAA 21-67-C-0041. Professor John O. Outwater was the Principal Investigator; he was assisted by Mr. William O. Carnes and Mr. Gerald Desany.

The work is being administered under the direction of the Plastics and Packaging Laboratory, Feltman Research Laboratory, Picatinny Arsenal, Dover, New Jersey, with Dr. Elise McAbee as Contract Project Officer.

This final report covers the period September 1, 1966 to August 31, 1967, and summarizes various phases of the work on this contract mentioned in more detail in the earlier quarterly reports. It shows in more detail the theoretical foundations that the work will be guided along during the next 12 month period.

The results presented in this report represent work in progress and may be subject to revision as the program continues.

ABSTRACT

A critical analysis of the sources of fracture energy in a composite material shows it to be dependent on the energy of debonding of a single fiber from the matrix material. This energy has been measured using a novel technique. A theoretical basis for the fracture energy of a composite has been developed relating it to the geometry, debonding energy and frictional force on an individual fiber. It shows that we can expect a more brittle composite if we reduce the fiber diameter, increase the bonding energy, increase the frictional force or reduce the fiber density in the laminate. These factors appear to be confirmed in actuality.

Measurements have also been made on the fracture velocity of a crack through resin showing it to be made up of a high velocity region and a low velocity region with substantially similar fracture energies. The actual crack velocity depends on the relative lengths of cracking time at the two different velocities.

On The Fracture Mechanics of Glass Reinforced Resin

The resin performs several functions in a laminate: it acts as a readily formable substance that becomes rigid, thereby giving the external shape to the item being produced; it protects the fibers from abrasion; it transmits external loads to the fibers and internal loads from fiber to fiber within the laminate; it may chemically interact with the finish or the substance of the fibers. It has, in fact, many functions that can be readily deduced from the very nature of the composite and which common sense rightly prescribes as its function. However, if we observe the difference in behavior on fracture between wood fibers embedded in an epoxy resin and those embedded so as to form the fibers of ordinary wood, we are immediately struck by a fundamental difference in behavior upon fracture and one which can be attributed to the nature of the matrix. The fibers embedded in resin snap in a brittle fashion and, though they improve the strength and modulus of the laminate, the material is brittle. The fibers in the wood confer strength and also have great energy absorbing qualities upon fracture.

There is apparently some real difference between the effects of the resin in each case and here, again, common sense comes to the rescue by indicating that the lack of bonding between the resin and the fiber may well give us the enhanced fracture energy characteristics of the wood as opposed to the fiber-resin block. If we explore this observation a little more deeply from the point of view of fracture mechanics we can obtain some unusual and important insights in regard to an important--perhaps vital--function of the resin which may be to permit some lack of bond between the resin and the fiber so that the fracture energy of the composite may be increased substantially above that of the components of the laminate. The fracture energy of glass is about 0.04 lb./in., that of resin 1.26 lb./in., and that of

reinforced plastics about 1000 lb./in. This latter approximation can be computed from data of Outwater (1). The difference between the three is so great as to suggest a different mechanism altogether in regard to the energy absorption within laminates during fracture. It is the purpose of this paper to formulate and demonstrate such a mechanism.

Fracture Mechanics

The science of fracture mechanics is concerned with the energy release rate upon fracture and implies that a crack in a material will extend catastrophically if the total energy required to extend a crack is less than the strain energy released in the material upon extension of that crack. With ideally brittle materials, where all the energy is assumed to be utilized in the creation of new surface, as in Griffith's original and well known hypothesis, the maximum stress that a material under plane stress circumstances could withstand is given by $\sigma = (2E\gamma/a)^{\frac{1}{2}}$ where σ is the stress, E the modulus, γ the surface energy and a the depth of the initial crack. This equation has the form $\sigma(a)^{\frac{1}{2}} = \text{constant}$ and it can be shown that this form works tolerably well not only with brittle materials, but also with ductile materials where the energy goes into the formation of fresh surface in only very small measure and by far the greater part of it goes into plastic deformation.

The symbol k represents the stress-intensity-factor and is a property of the material and of the method of loading and can be computed for the several forms of initial crack. In the Griffith case cited, $k = (2E\gamma/\pi)^{\frac{1}{2}}$. The crack extension force $G_I = (\pi/E)k^2$ for conditions of plane stress as might be found in a laminate during cracking. The energy absorbed then upon unit area of crack extension is G_I and it is reasonable to assume that the resistance to crack extension is the non-recoverable strain energy

loss regardless of where this lost energy goes. Using this understanding of G it can be shown that $G_I = \frac{P^2}{2}(dC/da)$ where P is the load on a member, C is its compliance and a the depth of a critical crack in that member upon onset of continuous cracking. If the crack-extension force is low, we can consider a material to be brittle, if large, then the material will require more energy to develop a crack and it would appear less brittle. It is an important property of materials and one that is of overriding importance with fiber reinforced laminates. The lack of brittleness of reinforced laminates is one of their important assets and one which, with the advent of new materials, we are finding can be taken less and less for granted.

The fracture energy of fiber reinforced laminates is peculiarly hard to measure on a meaningful manner and this is particularly true when we are dealing with glass reinforced laminates which have an extremely high value of fracture energy--so high, in fact, that it never has been measured satisfactorily. The meaning of the fracture energy is also open to question in such materials: they fail in a different fashion from most other materials.

Laminates are frequently found with "craze cracks" running through them that do not have any serious effect on their strength. These cracks are often present in great numbers and in close proximity to each other. They can in no way be considered a failure of the material though they certainly are a failure of the matrix and a visible crack in the laminate.

The definition of crack penetration which is often used to describe failure does not then apply in the case of laminates. If we define the depth of a failure crack as the depth to the last broken filament, then we are again in difficulties; filaments often debond from their embedment without breaking and then take much less load than the adhered filament. Their load carrying then is seriously reduced, but they are not broken. So

we again have to revise our definition of failure crack and the definition that we shall use is that the failed crack will be measured to the position of the last filament that has failed to debond and lie loosely within the matrix or has snapped inside or outside its matrix sheath. The analysis of this form of failure can be studied in two parts:

1. The instability and debonding of a filament within a matrix.
2. The relationship between this debonding of a filament with the fracture energy of the laminate.

The Measurement of Fracture Energy of a Resin Casting Without Fibers

One of the simplest methods of measuring the fracture energy of a material is by using a double torsion technique of Outwater and Gerry (2). This method can be readily adapted to resin. Samples of resin with composition--100 pts. EPON 826; 90 pts. NADIC Methyl Anhydride; 1 pt. DMBA--were cast and cured at 250°F for 24 hours before being tested. The results gave a fracture energy of pure resin of 1.26 lbs./in. This result corresponded well with other investigators but, on examination of the fracture surface, a series of lines were observed to cross the surface at the contour that the progressing crack front assumed, fig. 1.

Rather than investigate the energy for different formulations of resin, it was decided to observe the fundamental significance of these lines. It was quite apparent that the lines occurred where the crack hesitated before moving on again in traversing the whole sample. The nature of this hesitation and the forces involved were examined first by measuring the forces and hence deducing the fracture energies involved in cracking the specimen in the lined and unlined portions. The plot of the force against crack depth is shown in fig. 2. The lines across the surface

are apparently the points of where the crack slows down until the force builds up sufficiently to restart it again. The fracture energy can be measured accordingly both at the points of hesitation and as the crack moves rapidly. At the lines, it builds up to a maximum of 1.51 lbs./in. and, as the crack moves more rapidly, it falls off to a minimum value of 1.09 lbs./in.

The behavior of the crack front deserved careful study: the fracture energy varied from a maximum of 1.51 lbs./in. to a minimum of 1.09 lbs./in. and then the crack slowed down until the force built up again to start the crack off again at its high speed. The energy then depended on the speed of the crack. The measurement of the crack speed was then important and it was attempted by many methods. A high speed camera (4000 frames per second) was tried. This proved futile as the crack moved too rapidly for it to be reliably recorded similarly, difficulties were encountered using audio oscillations to vibrate the sample during the crack movement. The technique adopted was to use a 50,000 cps ultrasonic oscillator as in fig. 3. The crack surface now showed a series of lines as in fig. 4 and the spacing of these lines would show the actual crack velocity which could then be related to the fracture energy. The problem was not quite this simple as the lines where the crack hesitated was too slow moving for the ultrasonic ripples to be resolved so a 100 cps audio vibrator was used in this case. The ripples with the 100 cps oscillator is shown in fig. 5, and now a composite picture of the fracture energy at different velocities was obtainable as in fig. 6.

This plot of fracture energies as related to crack velocity is particularly interesting. It shows that there are two domains of velocities at which a crack can travel and that the same range of fracture energies is

used in both domains. The start-stop nature of cracks is cycling between these domains in the approximate path shown on the curve. The observed speed of a crack appears to be composed of a mix of the high speed portion and the low speed portion with the apparent speed being determined by the proportion of time spent at the higher speed compared to that at the lower.

An additional value that is vital in regard to the use of resins as structural materials is the fracture energy of the material under loading that is less than that needed to make a crack run. To find this value a torsional specimen was loaded at 80% of the value needed to make the crack run. The specimen was held at this load for a period of 9 months supported on a seismographic platform so there would be no disturbance at all. The result was that the crack did not move at all--zero movement. An immediate conclusion can be drawn from this observation: at 80% of the load needed to crack the epoxy, there is no increase in crack depth. This is at variance with the behavior of glass as a lower load, that is still above a certain minimum, will merely reduce the crack velocity.

The Measurement of Fracture Energy of a Resin Casting with Embedded Fibers

The problem of measuring the fracture energy of a resin casting with fibers embedded in it seems superficially simple--repeat the experiment above with fibers embedded in the resin. This produces numbers but has little significance. The fibers will tend to slow down the crack and produce a pattern as in fig. 7. The area covered, and the effect of the fibers is hard indeed to interpret and may, therefore, be considered invalid as a measurement method.

The next method tried was to use a specimen as devised by Mostovoy and Ripling (3) and shown in fig. 8. The fibers are cast into the specimen at

intervals and the double cantilever is pulled so that the crack will move observably down the center of the specimen and the cracking force recorded. The results of measuring the cracking force and plotting this with the depth of the crack is shown in fig. 9. The position of the fibers are directly related on the curve. The most obvious observation from this curve is that there is very little relationship indeed between the force and the location of the fibers. The reason for this is that the fibers pull out of the resin under a critical load and hence add an awkwardly unpredictable element to any simple calculation in regard to the fracture energy. Their contribution is not measurable from this experiment. A similar force distance plot without the fibers is shown in fig. 10 and the similarity and the singularities are apparent.

The function and the contribution of the fibers to the fracture energy and hence the effects of finish, length, etc. that we seek are not to be obtained from such an experiment. If we are to obtain these data we must use somewhat greater insight into the meaning of the fracture energy and into the behavior of the fibers. Particularly we must consider their pulling out and their debonding and the fact that the fibers themselves can make substantial contributions to the fracture energy without the resin being present at the crack tip--the crack, in fact, with reinforced laminates, is not at the crack tip of the resin but rather at the point of total failure or, as we shall see, at the point of pullout of the fibers. The mechanism of fiber debonding will first be considered.

The Stability of an Embedded Fiber

If we consider the energy factors involved in a cylindrical fiber that is embedded in an elastic matrix we can obtain an expression for the

spontaneous debonding of the fiber under a tensile load as shown in fig. 11 where a fiber has been pulled out of a cantilever specimen.

Consider an infinitely long fiber of diameter a and cross-section area A embedded in a semi-infinite solid as in fig. 12. Let P be the tensile load at the end of the fiber protruding a distance L from the surface and let the fiber debond a distance x into the matrix. Let G_{II} be the energy required to debond the fiber.

The strain energy in the filament is $\frac{P^2L}{2AE} + \int_0^x \frac{(P - \tau\pi ax)^2}{2AE} dx$, where τ is the shear force between the fiber and the resin after debonding. The energy required to debond an added length dx is $G_{II}\pi a dx$ where a is the diameter of the fiber. The strain energy released by an increment dx of debonding is $\frac{(P - \tau\pi ax)^2}{2AE} dx$. Equating these values we obtain:

$$\frac{(P - \tau\pi ax)^2}{2AE} = G_{II}\pi a$$

or, substituting for P we obtain:

$$(\sigma - 4\tau\frac{x}{a})^2 = (8E/a)G_{II}$$

where σ is the stress that the fiber bears at the end not embedded in the resin.

Using this equation we can predict several points of behavior of embedded fibers that may help to explain the behavior of laminates. Particularly, if the friction of τ is small, then the maximum stress that a fiber can bear before it pulls out will be inversely related to its diameter and will be independent of its length.

This simple formula, if it can be demonstrated experimentally, will enable us to determine two important parameters concerning the reinforcing of plastics:

1. The strain energy release rate on the debonding of a fiber within a resin matrix.
2. The frictional or other force gripping the fiber within the resin after the initial bond has been overcome.

It will ultimately enable us to relate the debonding strain energy release rate with surface finish, type of fiber, matrix, etc. and the frictional effect also can be related to these parameters. In order to check the formula a series of experiments were run where a monofilament was cast in a long resin specimen and this was bent in flexure after including a rubber release wall so that the specimen would crack and flex to pull out the fiber as in fig. 13.

This proved extremely difficult as the flexure bonding load of the rubber was great enough to throw serious doubts on the validity of any pull-out stresses. The method could, however, be modified by embedding the filament in a resin block, severing the fiber and then loading the fiber in compression and observing at what matrix stress the fiber would debond.

A diagram of the method is shown in fig. 14. In this case the stress in the fiber at a depth x, if the filament has debonded to that depth, is $4\tau x/a$. The strain in the filament before debonding is the same as that in the resin or ε_r. The stress in the filament then is $\sigma_r E_f/E_r$ where σ_r is the stress in the resin. The loss of strain energy in the filament upon debonding a distance dx is:

$$\left\{\left(\sigma_r \frac{E_f}{E_r}\right)^2 - \left(\frac{4\tau x}{a}\right)^2\right\} \frac{\pi a^2}{8E_f} dx$$

and this must equal the product of the strain energy release rate and the area debonded or $G_{II}\pi a dx$ where

$$G_{II} = \left[\left(\sigma_r \cdot \frac{E_f}{E_r}\right)^2 - \left(\frac{4\tau x}{a}\right)^2\right]\frac{a}{8E_f}$$

This formula operates in compression and on a simple specimen. It gives not only the value of G_{II} which can be readily measured when the debonding length is zero, but also the change in stress on the resin necessary to force the debonded length to greater values which will give us a value of the friction between the filament and the resin. This value has been determined using the arrangement shown in fig. 14 and gives values of G_{II} for the filament treated with A-1100 as 1.14 lb./in. This value is an average of fifteen specimens.

This method will be used to show the values of G_{II} and of the frictional shear stress between the resin and the fiber for different treatments and cure conditions and will greatly help in our understanding the nature of bonding in the reinforcement process.

The understanding of the energy absorbing mechanisms in the filament can be directly extended to the laminate itself.

The Fracture Energy of Resin with Fibers Embedded in it

As mentioned above, the fracture energy of fiber reinforced resin is far greater than that of the resin alone. There is often a situation where "craze cracks," or the pre-existence of cracks passing through the resin without affecting the strength of the laminate. The fracture energy of the resin will certainly be an insignificant contribution to that of the composite in the second case, and it will also be a very small portion of the whole in the first case; so we will be correct in seeking the reason for the high values of fracture energy in a laminate as being substantially independent of the resin fracture energy. The failure crack releases

energy from the resin component of the system and this is normally much less than that of the whole.

Let us now apply fracture mechanics concepts to the development of a crack passing through the fibers connecting two halves of a rigid member such as might occur on a laminate that has suffered a craze crack. Such a situation can be idealized by considering a number n of similar filaments of cross-sectional area A, each of unloaded length y stretched between two rigid members C and D unyieldingly embedded in each member as in fig. 15. The two members would be pulled apart with a load P and an increment of load δP would produce an increment of extension δy then $\delta y = \delta P\, y/nAE$ where E is the modulus of the fiber.

The compliance C of the system is given by $\delta y = C\, \delta P$ then $C = y/nAE$, but if we consider a crack of depth z entering from the left and consider the whole specimen to have an initial length b and width w and containing originally n_o filaments then the area fraction of the specimen devoted to fiber substance will be:

$$A_f = n_o A/bw$$

and the actual number of fibers bearing the load after a crack penetration z will be given by:

$$n = (b - z)n_o/b$$

so,
$$dn = -n_o dz/b$$

The problem of determining the fracture energy of a laminate then resolves itself into finding the appropriate value of y which would be the debonded length of the filament during the loading process of the laminate. It is the energy released from this debonded portion that is in excess of that released from the fiber substance and resin substance alone that is

responsible for the large value of fracture energy associated with reinforced laminates compared to that of resin or fiber substance alone.

The debonded length y/2 can be computed as follows: based on fig. 12, let σ_r, σ_{f1}, E_r, E_f be the stresses and moduli of the resin and fiber within the resin at the edge of the debonded portion. The strain in the resin and in the fiber will be the same within the resin where there is no debonding so:

$$\sigma_{f1}/E_f = \sigma_r/E_r$$

But if A_f and A_r are the areas of fiber and glass in unit area of laminate then:

$$A_f \sigma_f = A_f \sigma_{f1} + A_r \sigma_r$$

or,

$$\sigma_f - \sigma_{f1} = \sigma_f/\left(1 + \frac{E_f A_f}{E_r A_r}\right)$$

but, we considered the pulling out of a single strand, we found:

$$(\sigma_f - 4\tau x/a)^2 = 8G_{II}E_f/a$$

The modification to this equation demanded by the presence of the laminate compared to the filament alone pulling out is the substitution of $\sigma_f - \sigma_{f1}$ for σ_f and y/2 for x. We then obtain:

$$y = \frac{a}{4\tau}\left[\frac{\sigma_f}{1 + \frac{E_f A_f}{E_r A_r}} - \sqrt{\frac{8G_{II}E_f}{a}}\right]$$

This value of y gives us a strain energy release rate of:

$$\frac{\sigma_f A_f a}{2E_f \tau}\left[\frac{\sigma_f}{1 + \frac{E_f A_f}{E_r A_r}} - \sqrt{\frac{8G_{II}E_f}{a}}\right] \qquad (1)$$

This formula will apply when the fiber debonds within the resin matrix as a crack passes through it and in the presence of a craze crack along the path that the failure crack will follow. If there is no craze crack present,

then the value of the fracture energy will be the above expression plus the fracture energy of the resin alone.

It is interesting to note that there will be no debonding as the crack develops through the resin in the case when $(G_{II}E_f/a)^{\frac{1}{2}}$ is greater than $\sigma_f(1 + E_fA_f/E_rA_r)$; in this case the crack will cut cleanly through the resin and the fibers and the energy of fracture will be $G_{Ir}A_r + G_{If}A_f$. This latter case occurs when the bonding energy between the fiber and the matrix is very large such as with cotton threads in resin.

Equation 1 is important as it not only enables us to determine the actual value of the strain energy release rate upon fracture of a fiber reinforced laminate, but also shows how the various properties of the component parts contribute to the fracture properties of a composite material. It predicts that the smaller the diameters of the filaments or the higher the bond strength between the filaments and the matrix the more brittle the material. High modulus fibers will lead to a brittle composite, a more flexible matrix will also, and so will a reduced fiber content. These factors have not been apparent in the past as we have been so caught up with using glass-resin combinations that we have omitted examination of several of the fundamental interactions. An understanding of these will help explain some of the phenomena that we are observing with whisker reinforced fibers and with metal matrices. They are all proving brittle and the reason is observable from equation 1--the diameters are small and the shear between the filaments and the matrix is too good. So if we wish to improve the properties in regard to fracture energy, we must reduce the bonding and increase the diameter.

It should be noticed that one of the great advantages that we have enjoyed with glass fiber reinforced laminates is that the bonding is in

fact poor. Efforts have been made to improve it, but have not succeeded; had they succeeded, then the material would be much more brittle and perhaps less serviceable than it is. This is also true with the use of filaments with corrugated surfaces. If we use such filaments, then the effective bonding, in that the corrugations would interfere with slip between the fiber and matrix, would be greater and hence the material more brittle. It should also be noted that the actual pullout length of glass filaments is high as there is very poor bonding indeed at the end of a roving. The pull-out also is high with a monofilament but, as we can expect, the length-diameter ratio of the pulled-out length to be the same, the pullout length will, in fact, be much smaller for the finer fibers.

Conclusions

An examination of the possible mechanism of energy absorption in composite laminates during unidirectional stressing indicates that the lack of bond between the matrix and the fiber may be a vital factor in avoiding brittleness of the laminate. Computations based on this hypothesis closely approximate values measured for the crack extension force of glass reinforced epoxy resins.

This hypothesis further suggests that a reduction in the diameter of the fibers, a decrease of the modulus of the resin, an increase of fiber modulus, a decrease in fiber content or a reduction of fiber strength would all tend to increase the brittleness of the composite.

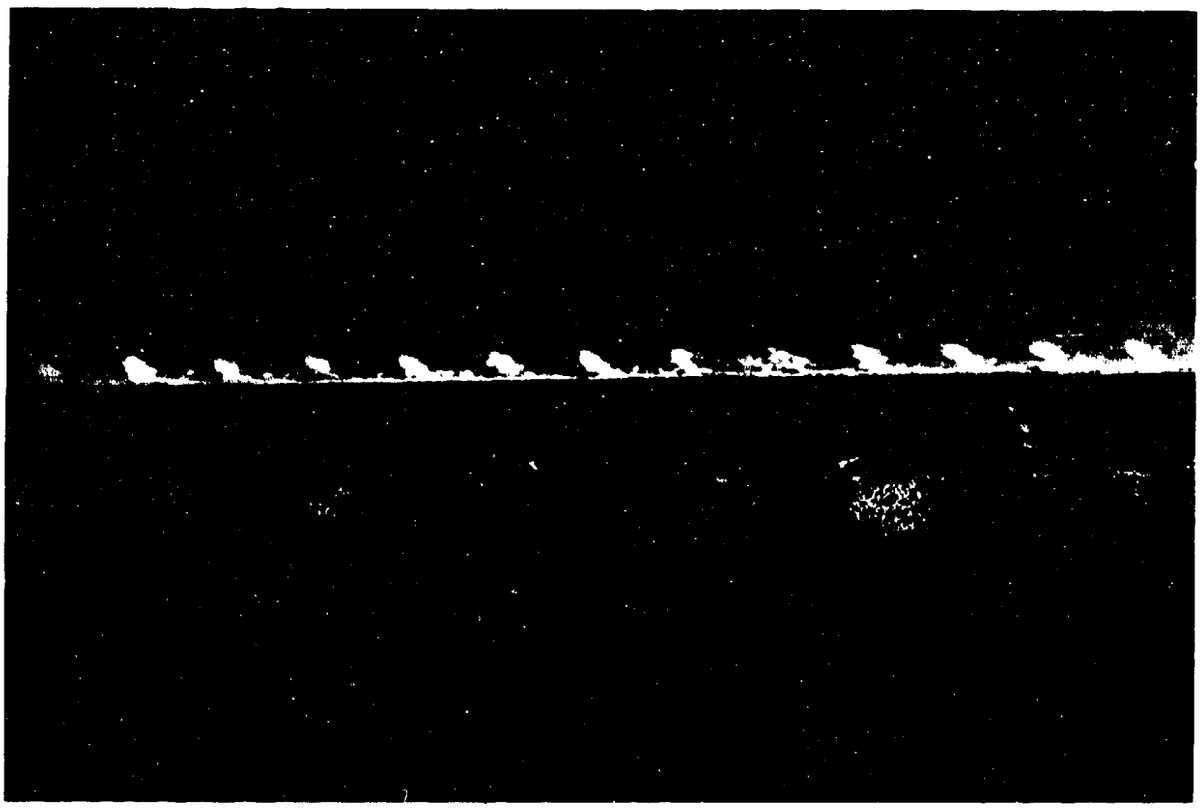

Fig. 1. Photograph of crack in resin showing points of hesitation of the crack front.

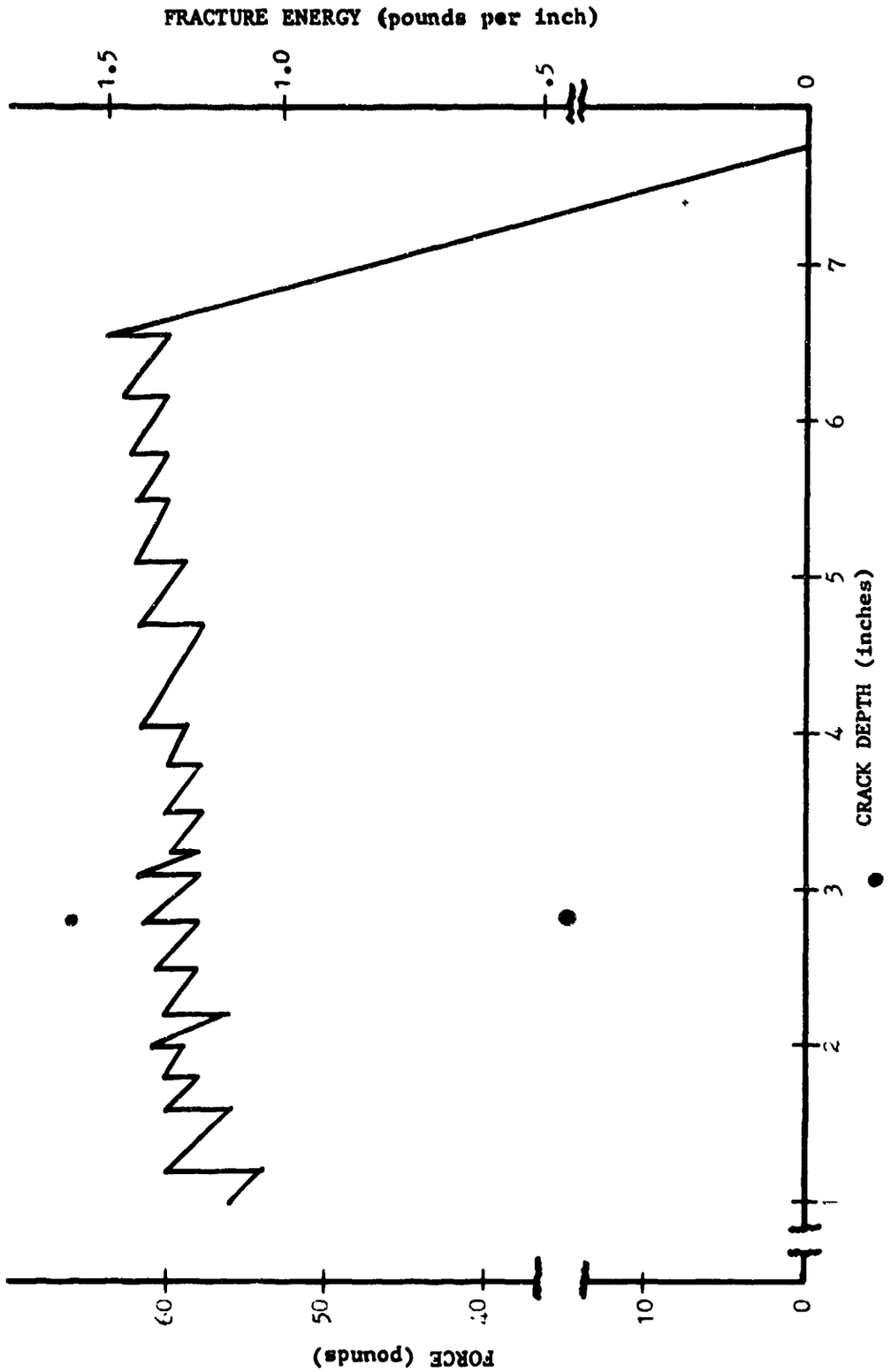

Fig. 2. Plot of fracture energy and force of loading vs. crack depth on double-torsion specimen.

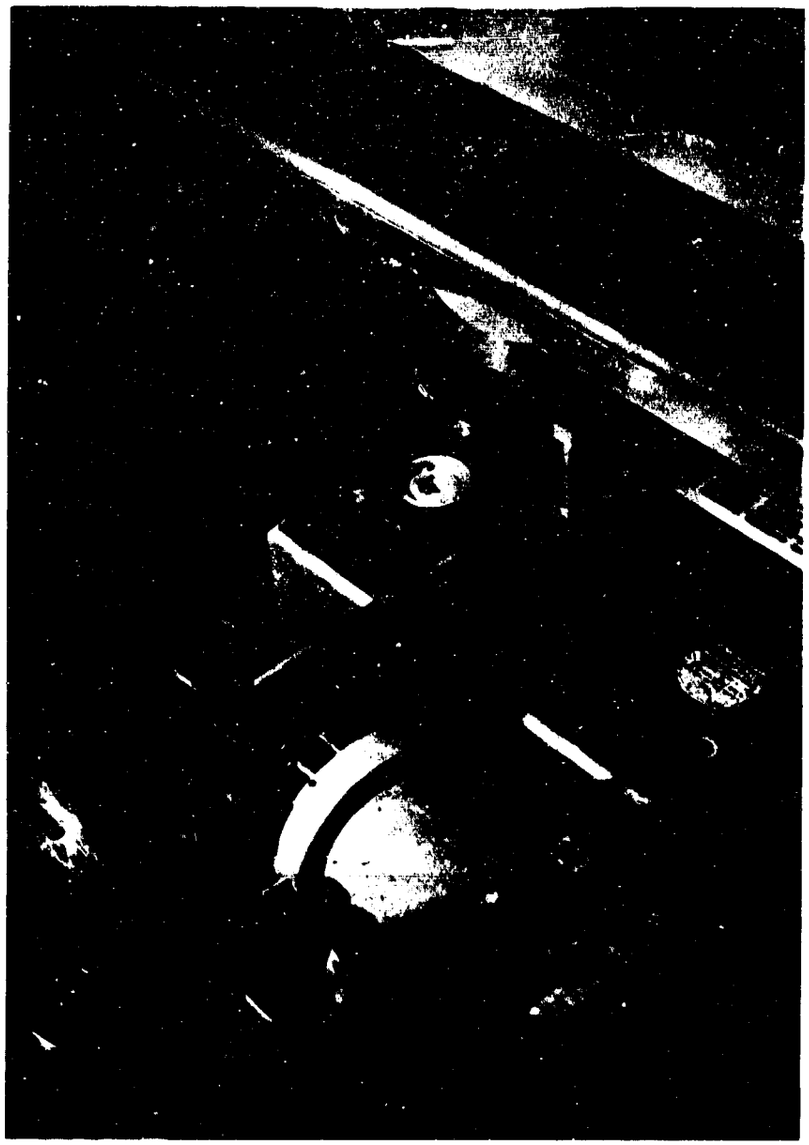

Fig. 3. Specimen being vibrated ultrasonically at 50,000 cps.

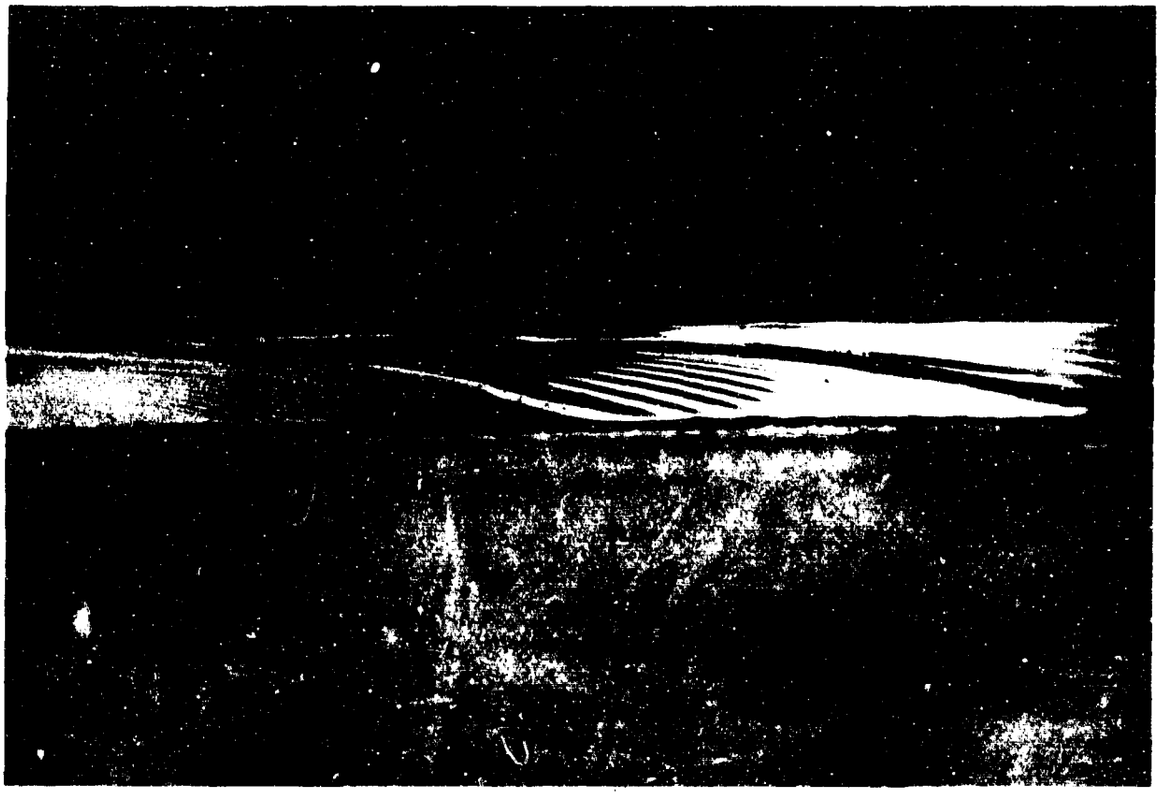

Fig. 4. Typical view of crack surface after crack was developed while the specimen was being vibrated at 50,000 cps.

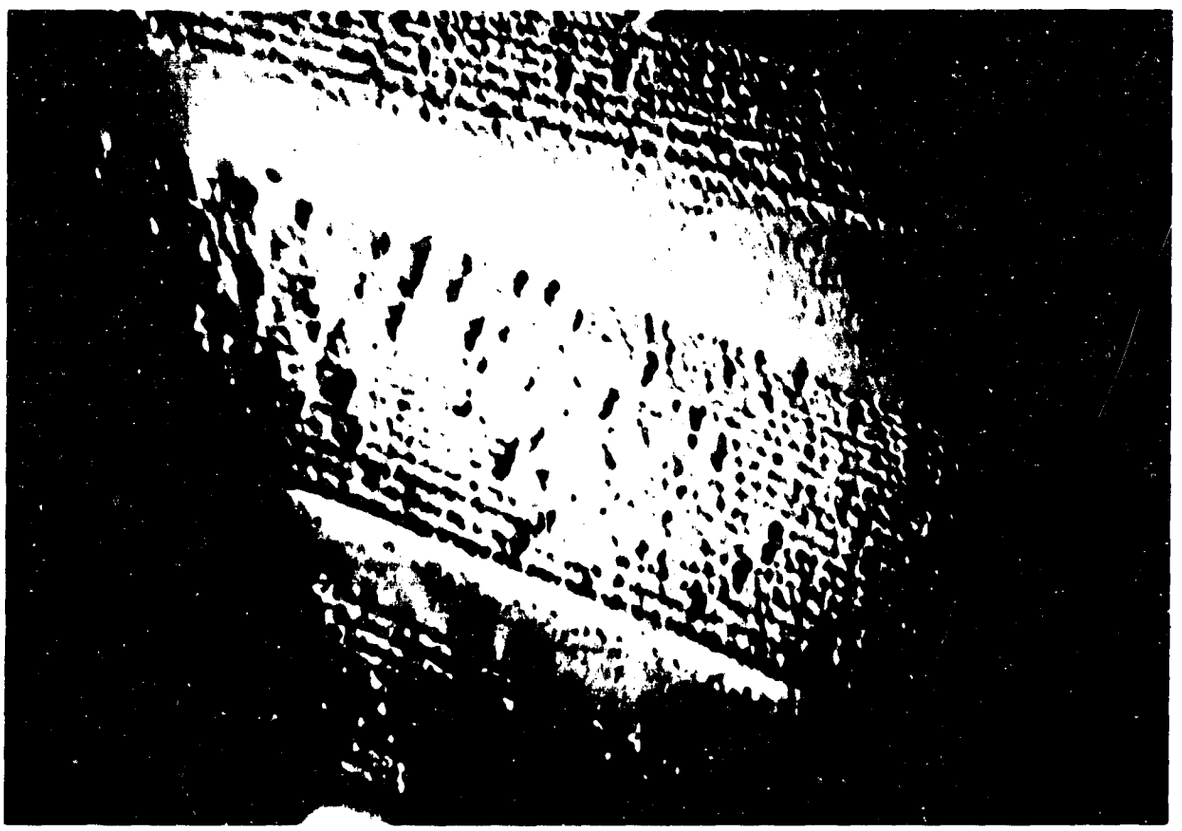

Fig. 5. Typical view of slow crack developed while specimen was being vibrated.

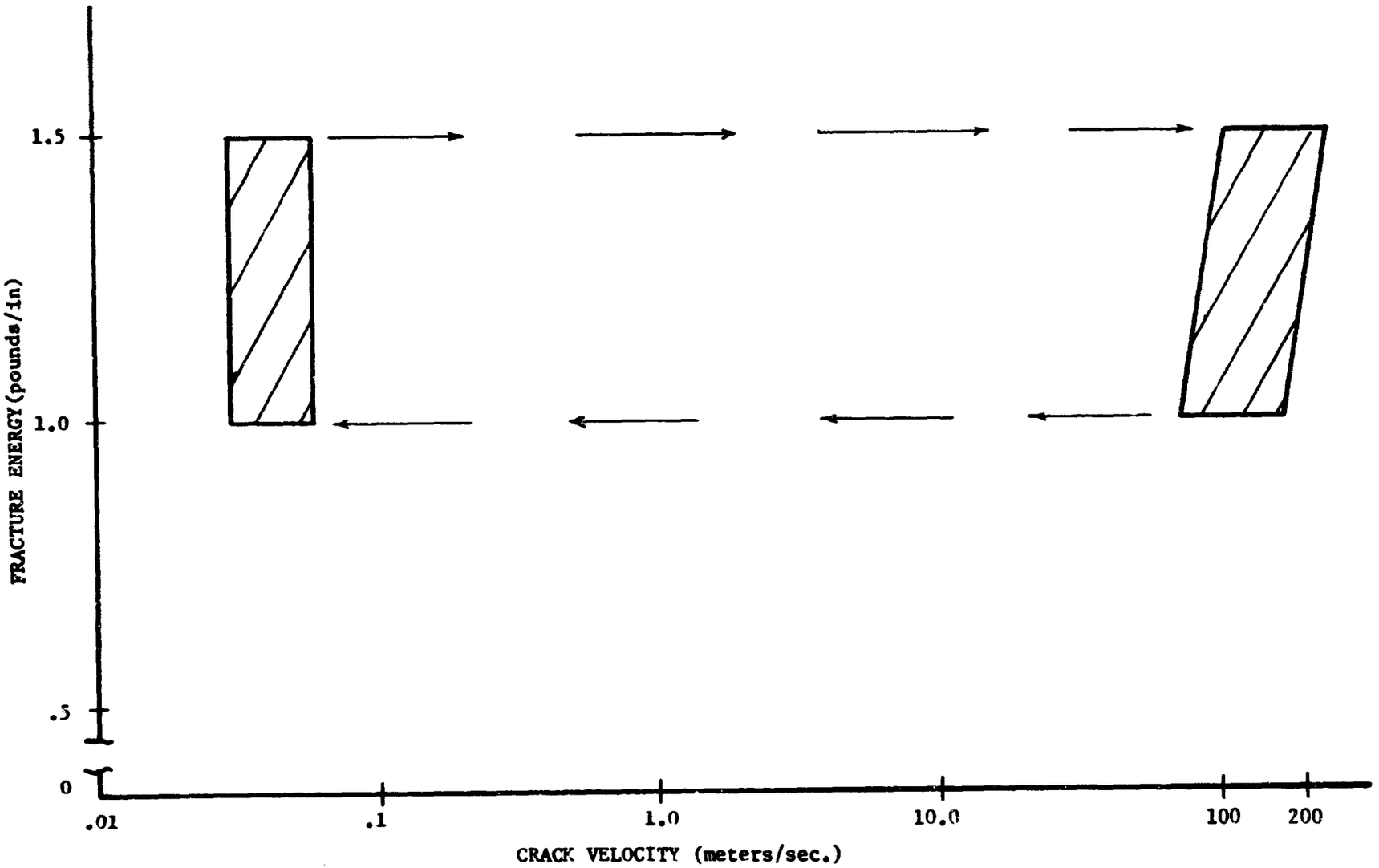

Fig. 6 Fracture energy vs. crack velocity during cracking of epoxy resin, showing energy-velocity path.

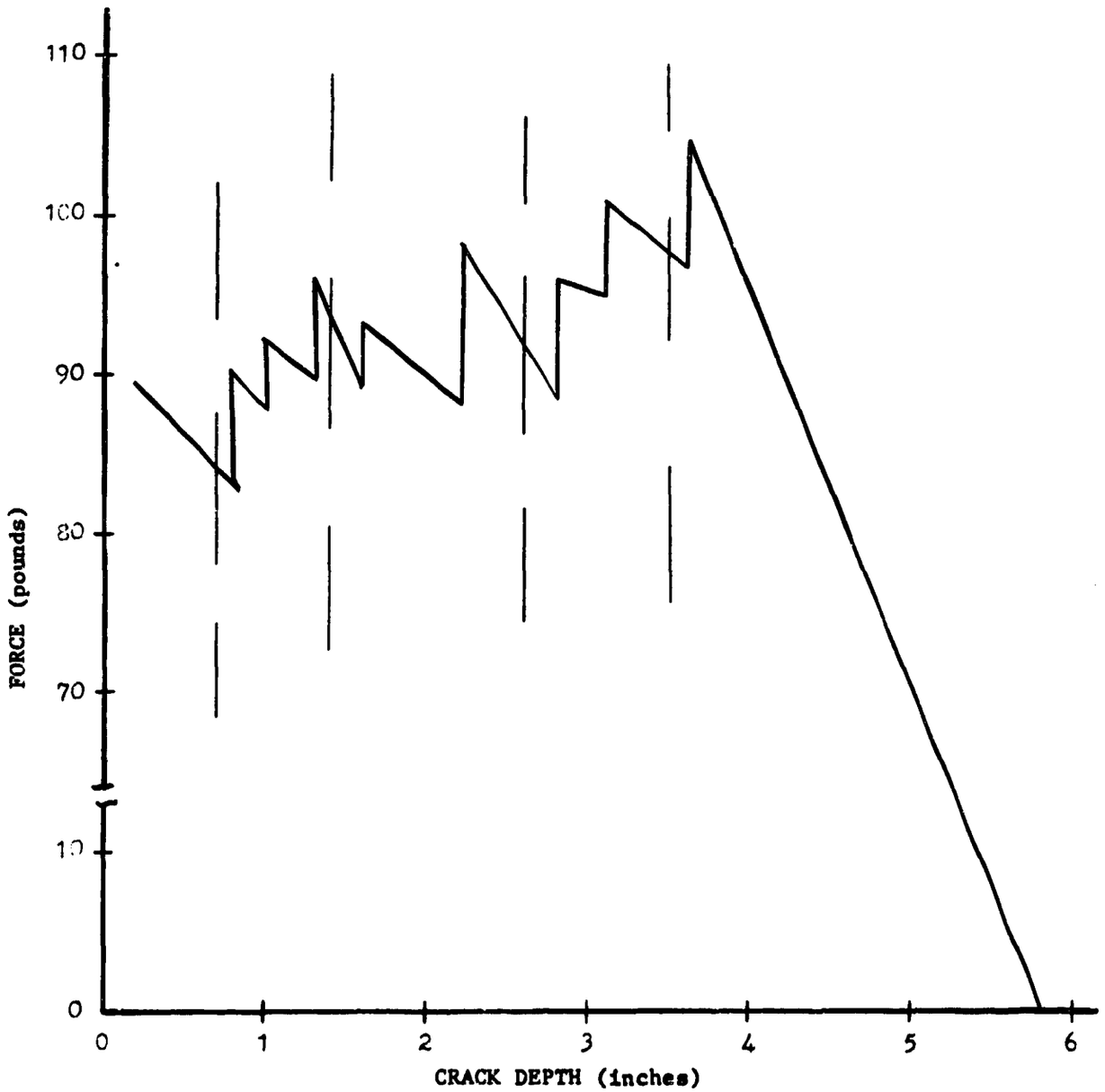

Fig. 7. Plot of force of loading vs. crack depth for a torsional specimen containing embedded fibers as shown.

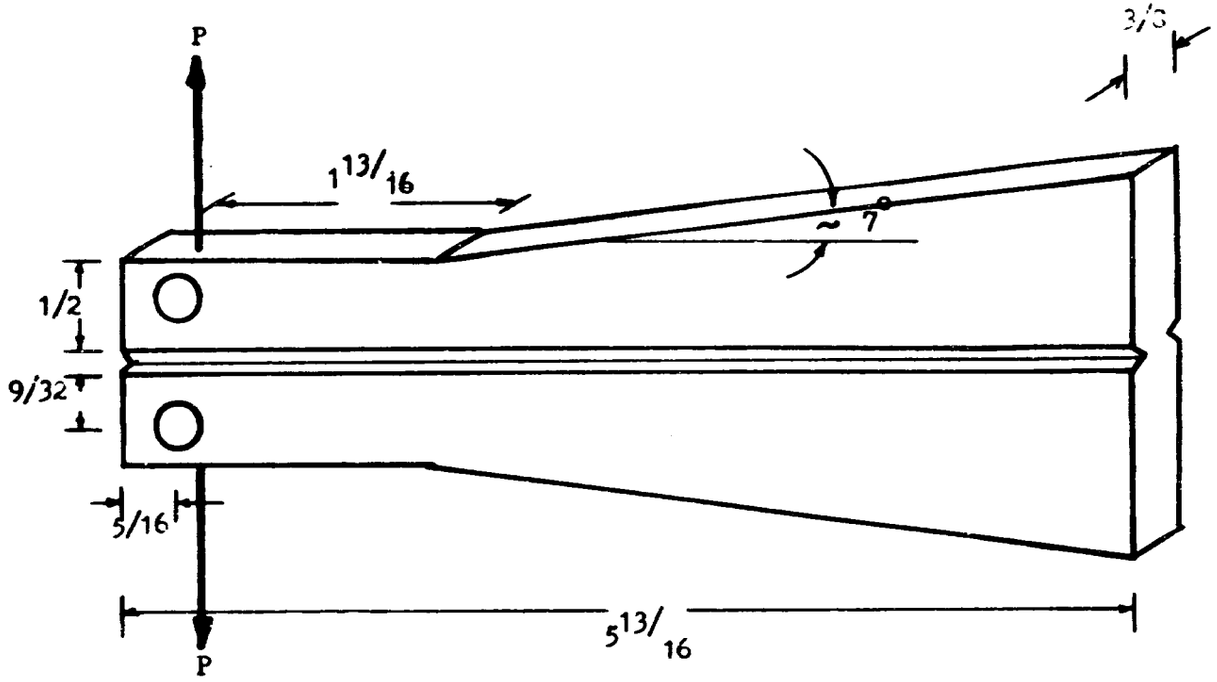

Fig. 8. Typical specimen using double-cantilever technique.

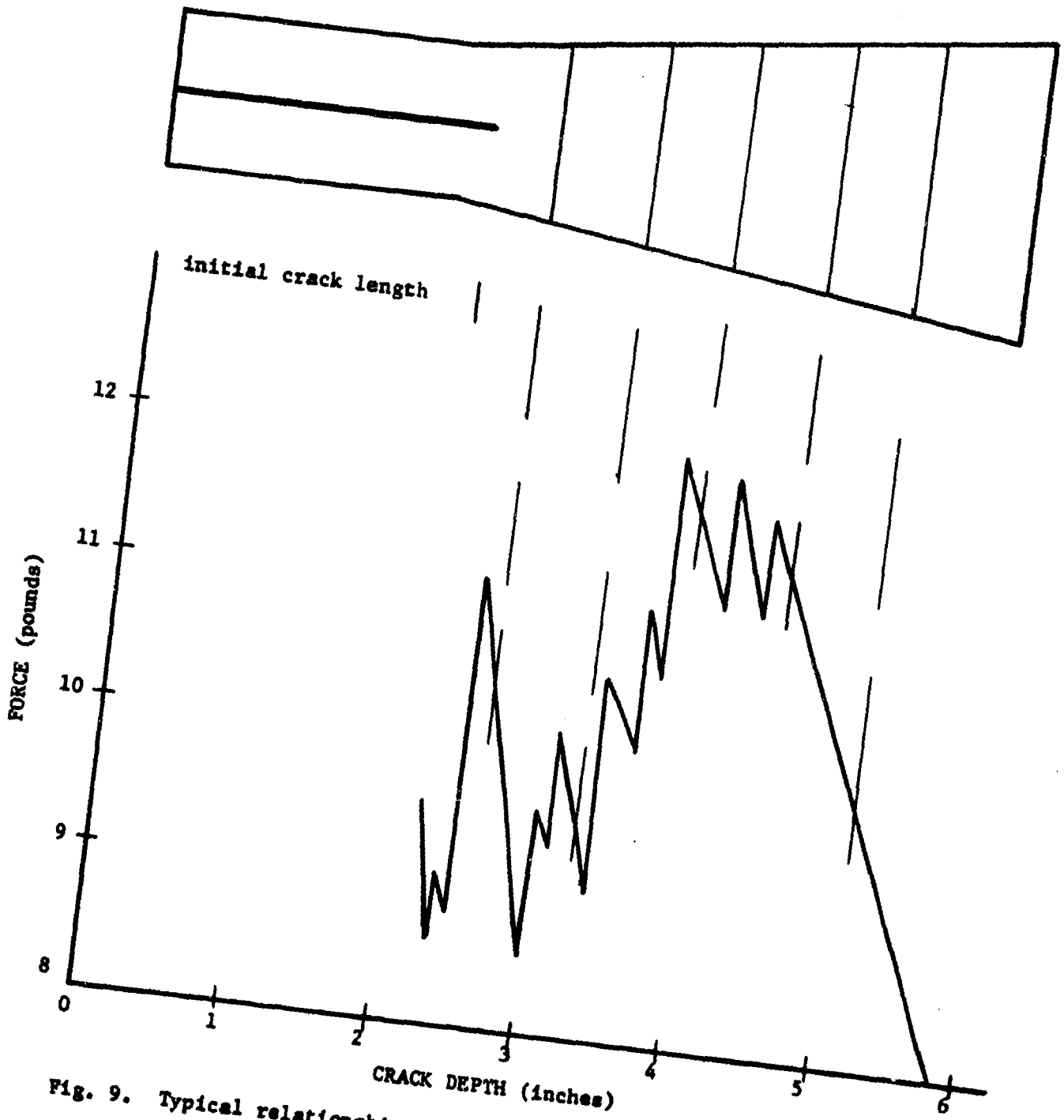

Fig. 9. Typical relationship between force and crack depth in double-cantilever specimen with embedded filaments in positions shown.

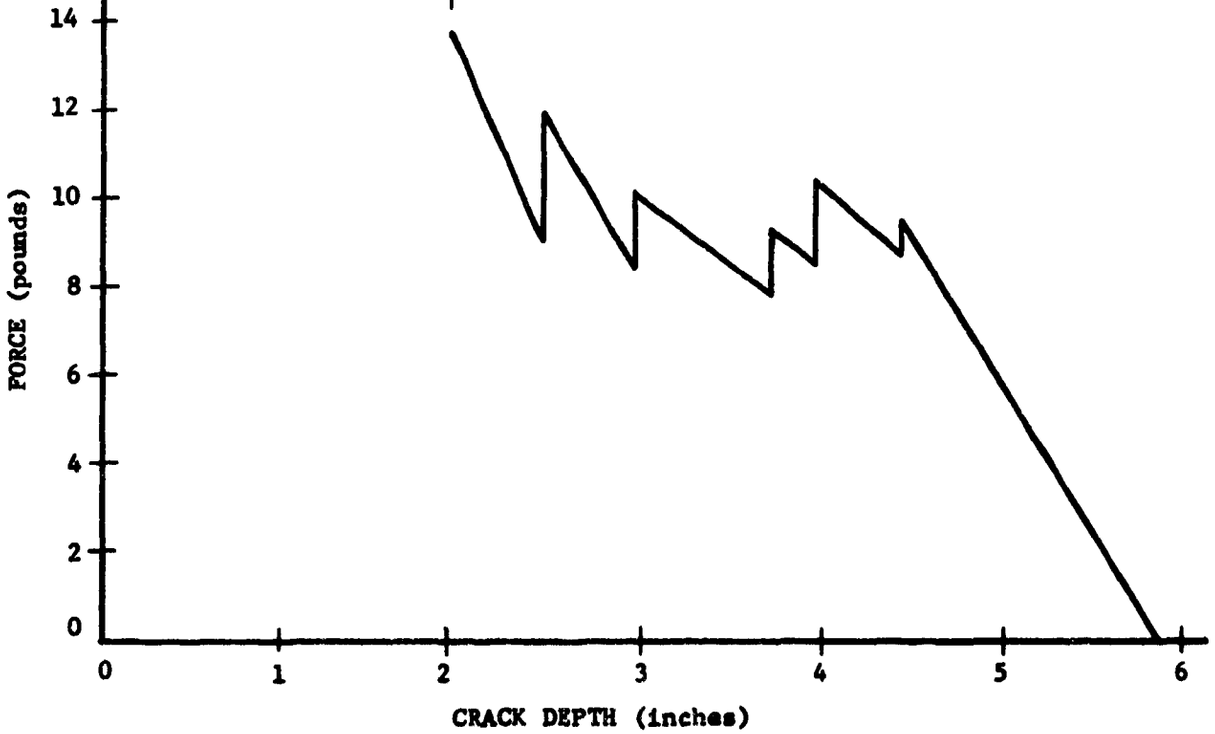

Fig. 10. Typical relationship between force and crack depth in double-cantilever specimen without embedded fibers.

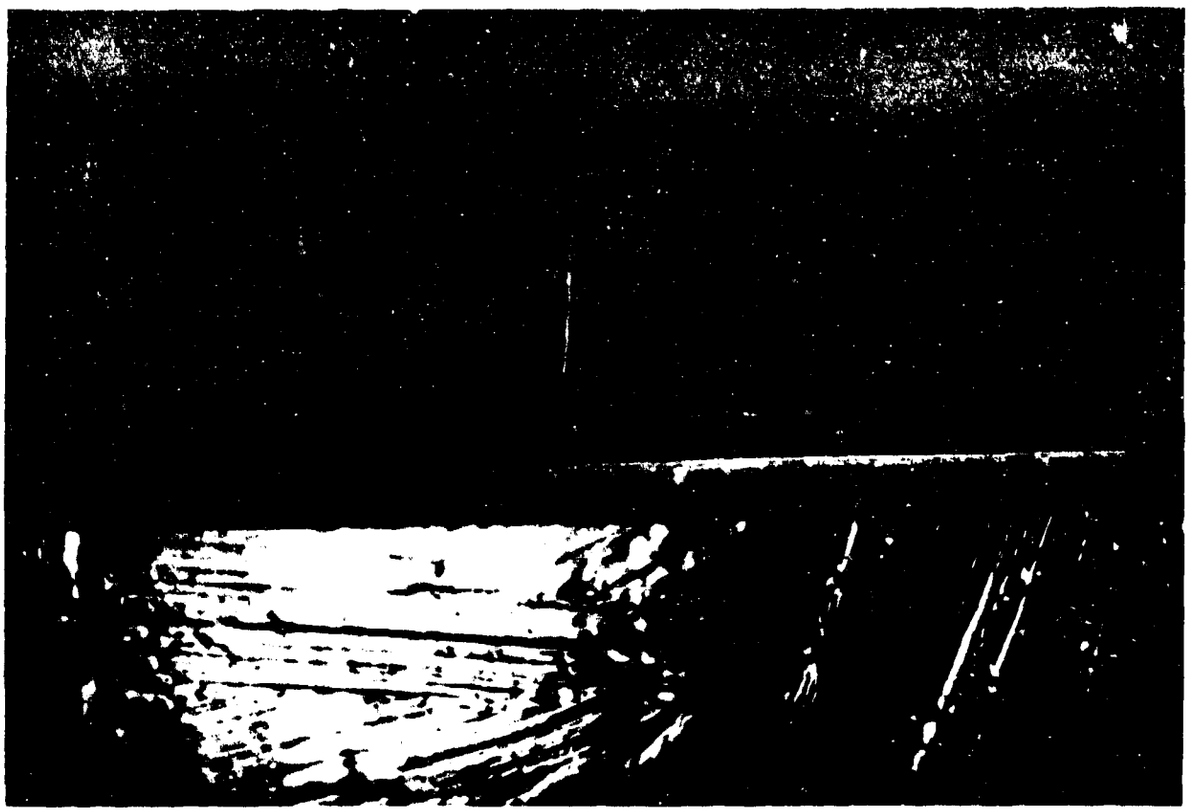

Fig. 11. Photograph of a double-cantilever specimen showing typical fiber pullout.

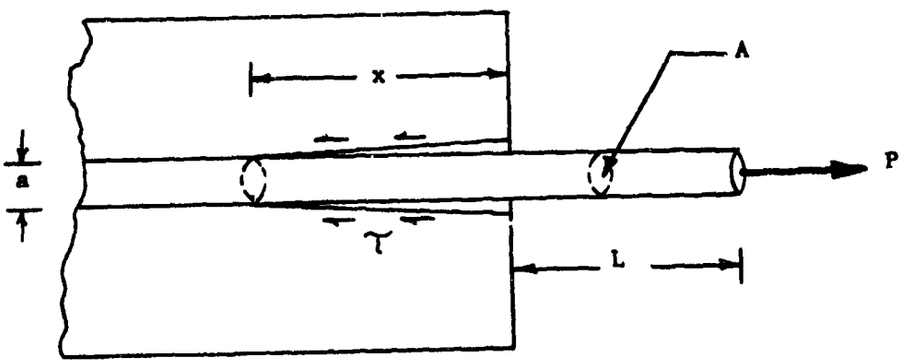

Fig. 12. Sketch of a single embedded fiber under tensile load.

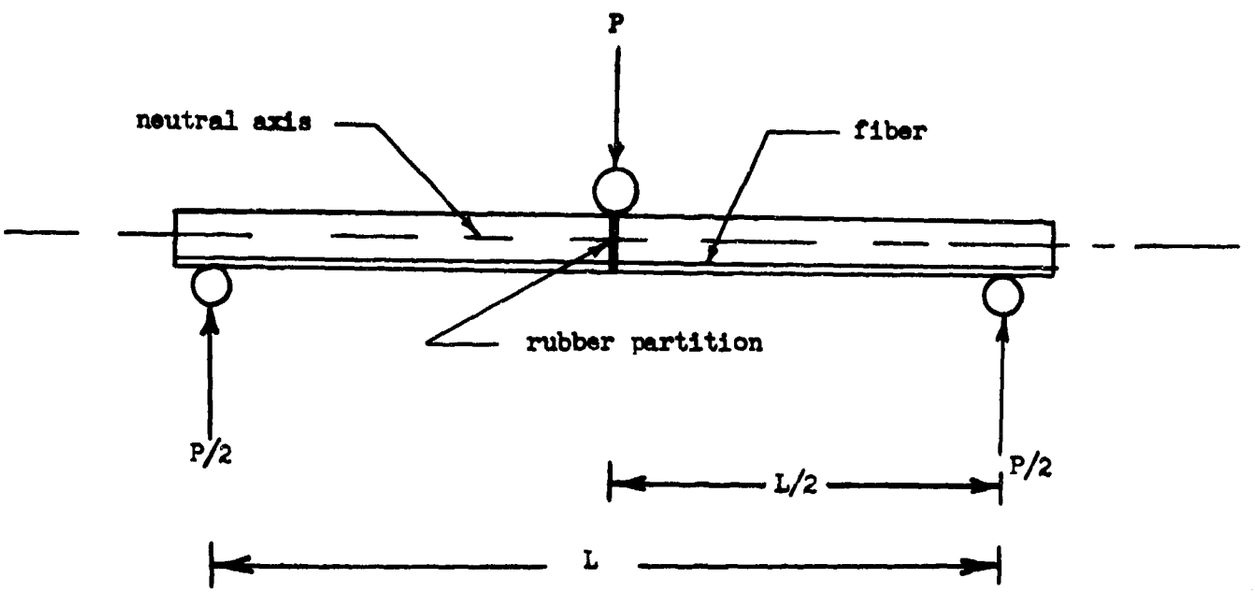

Fig. 13. Sketch of technique tried to determine the values of G_{II} of debonding between the fiber and the resin.

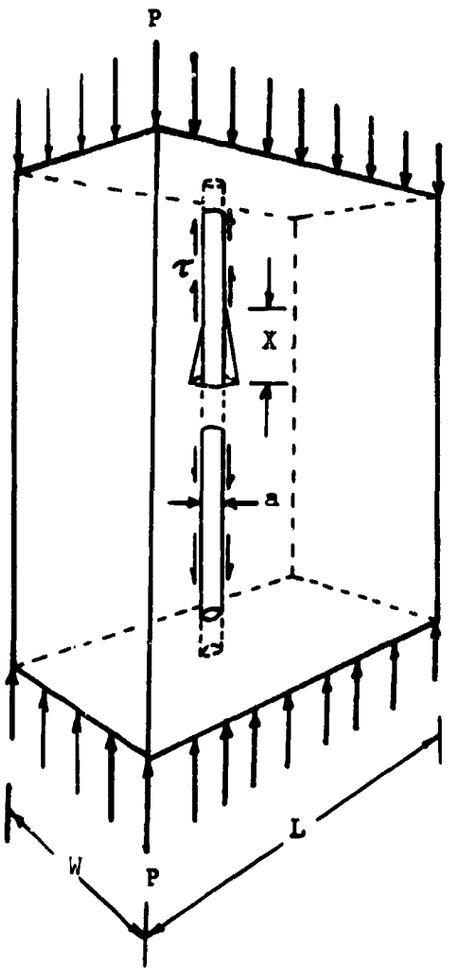

Fig. 14. Sketch of technique used to determine the debonding fracture energy between the resin and the fiber.

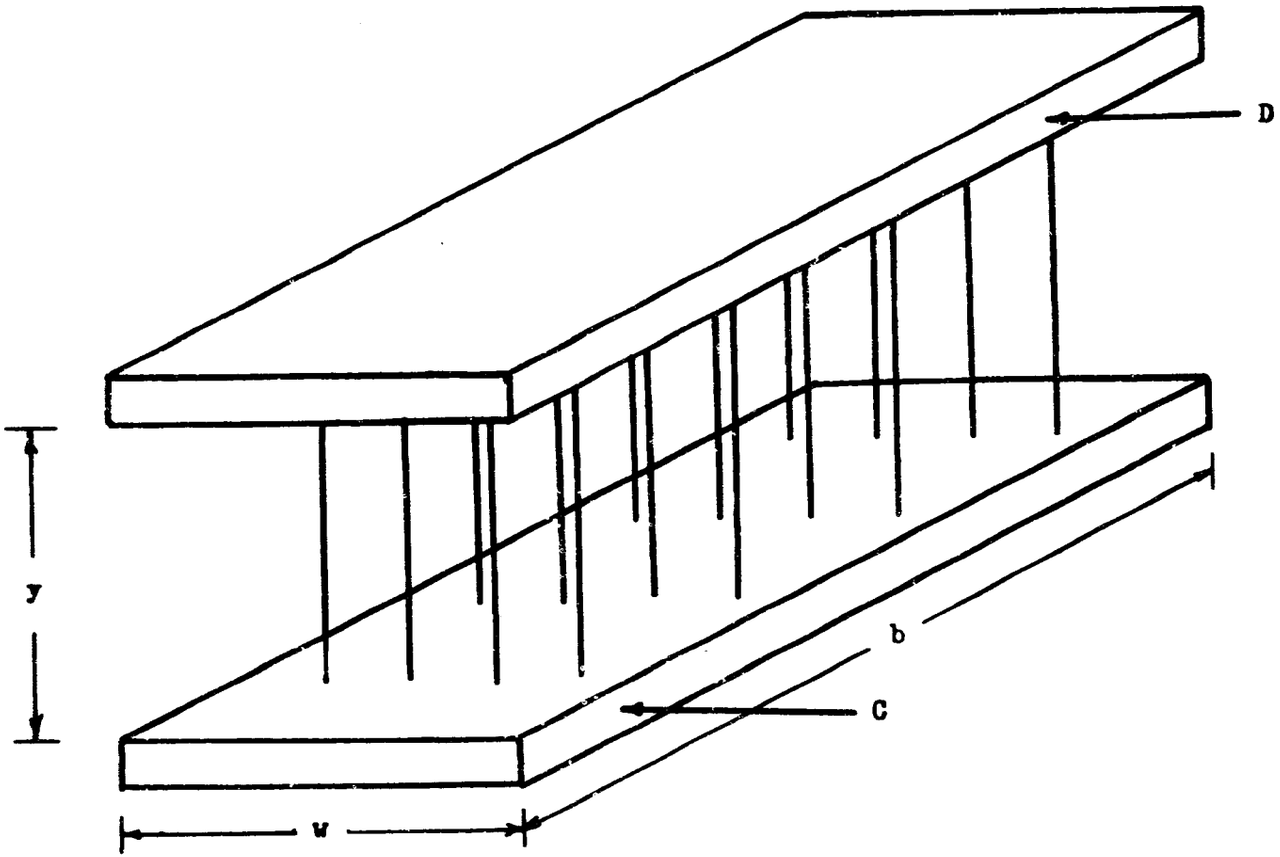

Fig. 15. Diagram of a unidirectional laminate with a debonded length of fibers y.

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13. ABSTRACT

A critical analysis of the sources of fracture energy in a composite material shows it to be dependent on the energy of debonding of a single fiber from the matrix material. This energy has been measured using a novel technique. A theoretical basis for the fracture energy of a composite has been developed relating it to the geometry, debonding energy and frictional force on an individual fiber. It shows that we can expect a more brittle composite if we reduce the fiber diameter, increase the bonding energy, increase the frictional force, or reduce the fiber density in the laminate. These factors appear to be confirmed in actuality.

Measurements have also been made on the fracture velocity of a crack through resin showing it to be made up of a high velocity region and a low velocity region with substantially similar fracture energies. The actual crack velocity depends on the relative lengths of cracking time at the two different velocities.

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